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Failure crack orientation at ductile shear fracture of $Fe_{80-x}Ni_xB_{20}$ metallic glass ribbons

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Orientation of failure cracks of oblique mode at ductile shear fracture are measured for $Fe_{80-x}Ni_xB_{20}$ (x = 10, 20, 30, 40, 50, 60) metallic glass ribbons (MGR) uniaxially extended with strain rate $\dot{\varepsilon}$ from 3.3×10^{-6} to $1.25 \times 10^{-3}\,s^{-1}$ at temperatures 300, 77 and 4.2 K. An angle between the failure crack and tension axis is found to depend non-monotonously on nickel concentration x and strain rate $\dot{\varepsilon}$. These dependences are probably due to the presence of dilatancy in shear bands and its variation as a function of concentration, strain rate and temperature.

Mean values of quenching stresses in ribbons are estimated by comparing experimentally measured values and those predicted by T. Thomas in 1953.

1. Introduction

High local plasticity of material is a characteristic feature of metallic glasses. Their plastic deformation is as high as several tens per cent at rolling, compressing and bending. Plastic deformation of metallic glasses at low temperatures ($T < 0.7\,Tg$, the latter being the vitrification temperature) occurs non-uniformly, through the formation of localized shear bands, whose width does not exceed $0.05\,\mu m$ while the local shear reaches 10 [1]. The appearance of such shear bands at uniaxial tension of MGR leads to the formation of rather high steps. These bring about local stress concentration which eventually results in ductile shear fracture of MGR.

Two modes of localized shear and ductile shear fracture have been reported [2]. The "45-degree" mode is characterized by the failure crack orientation along the plane of maximum shear stress, inclined to the broad face of a ribbon ("transverse" shear). In the "54-degree" mode the crack propagates along the plane that is parallel to zero elongation direction ("oblique" shear).

"Transverse" shear is associated usually with antiplane strain, arising near the front of a crack oriented perpendicularly to the direction of tension. It was supposed [2] that such shear takes place in wide ribbons (with the width-to-thickness ratio not less than 8), when the grip effect initiates tearing of the ribbon edge and causes the antiplane strain state. On the other hand, the "oblique" shear was considered as typical of spontaneously originated localized strain in a thin narrow ribbon, under uniform uniaxial tension, when the width-to-thickness ratio was smaller than 8.

According to the mathematical theory of plasticity [3, 4], the orientation of localized shear band (and hence of the failure crack at the "oblique" mode) is defined by purely geometrical considerations, that must provide for the absence of stress jump at the shear band boundary. At "oblique" shear in the zero elongation direction, the following conditions are satisfied: (a) Material continuity at the boundary of localized shear band; (b) coincidence of principal stress and strain axes, and hence of shear rate components, both outside and inside the band [3].

With due regard for these conditions, the theoretical analysis [3] predicts the values of the angle α between the shear band and the tension axis. These values of α are listed in Table I for the two criteria of plasticity and two values of Poisson's ratio: $\nu = 0.5$ (incompressible body) and $\nu = 0.35$ (metallic glass $Fe_{50}Ni_{30}B_{20}$) [5]. They are found [4] under the condition of constant volume in the course of plastic shear, i.e. at the absence of dilatancy. In what follows we will assume, in accordance with the results of [6], that plastic deformation in metallic glasses obeys the Tresca criterion.

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TABLE I Theoretical values of the angle α between the shear band and the tension axis [4].

ν	Von Mises criterion	Tresca criterion
0.5	54.7°	54.7°
	.	39.2°
0.35	59.4°	38.4°
	51.5°	

The concept of relative volume change accompanying a simple shear, i.e. of dilatancy, was introduced into the plasticity theory already by Reynolds [7]. Bowden and Jukes [8] showed that the angle α should change when a plastic shear is accompanied by dilatancy. This is caused by additional displacement of the shear band boundary due to the change of band volume which violates the conditions of zero elongation. To meet these conditions, the angle α should be increased accordingly, the higher the dilatancy [8].

In this work the failure crack orientation is studied in the oblique mode of ductile shear fracture at uniaxially MGR tension and the obtained regularities are compared with the existing state of art.

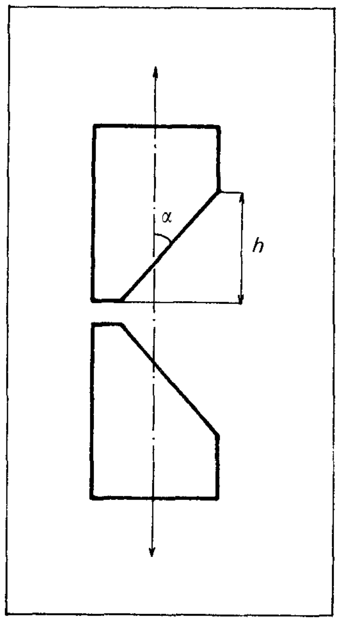

Figure 1 Schematic view of oblique–mode ductile shear fracture of MGR. α is the angle between the failure crack direction and tension axis; h is the fracture "height".

2. Experimental procedure

Samples of MGR $Fe_{80-x}Ni_xB_{20}$ (x = 10, 20, 30, 40, 50, 60) prepared by planar flow casting had the dimensions about $0.025 \times 10 \times 20\,mm^3$. They were tensioned with rates from $0.004\,mm\,min^{-1}$ to $1.5\,mm\,min^{-1}$ up to fracture at 300, 77 and 4.2 K. Fractographic analysis of fracture surfaces was accomplished in a scanning electron microscope "TESLA BS-300". The angles α was measured to within 1°, and the oblique fracture height $h = d.\mathrm{ctg}\,\alpha$, d being the width of oblique fracture zone, to within 0.2 mm (Fig. 1).

3. Results and discussion

3.1. Realization of oblique fracture mode

L. Davis [2] has shown that arising of oblique shear in wide ribbons is impeded by the twisting influence of the grips which leads to tearing of a ribbon edge and to arising of a shear in a one of the maximum shear stress planes ("45-degree" mode).

In this connection we used such a method of fixing ribbon samples in grips, that provided uniform uniaxial loading during the tension, and as the thing turned out, both oblique shear and oblique fracture mode became available in wide ribbons. As a rule these were usually observed as a combination of oblique and transverse fracture modes, schematically shown in Fig. 1. Electronmicroscopic observations showed a vein pattern on oblique mode fracture surfaces, being an indication of preceding "54-degree" shear.

3.2. Composition dependence of oblique mode orientation

Initial qualitative observations showed that the angle α is often very much different from 54°. Since the mathematical plasticity theory strictly indicates the dependence of α on material properties, in particular, on its compressibility [4] and dilatancy value [8], we conducted systematical measurements of α for different amorphous alloys under different strain conditions. These measurements have been carried out for the parts of fracture surfaces corresponding to the oblique shear. These parts were selected on the basis of microscopic observations.

To reveal the qualitative regularity preliminary measurements were not made of the α, but of the height h (Fig. 1). The averaged h values under various deformation conditions were found from four measurements. The results are shown in Fig. 2. The outlined correlation field indicates that h tends to maximum at x = 30 at % Ni.

We have studied the results of α measurements by statistical dispersion analysis to make clear how these results are affected by some factors of the experiment (Ni-concentration x, $\dot{\varepsilon}$ and T). In Table II average values of α are listed for different nickel concentrations, strain rates and temperatures. 90% intervals

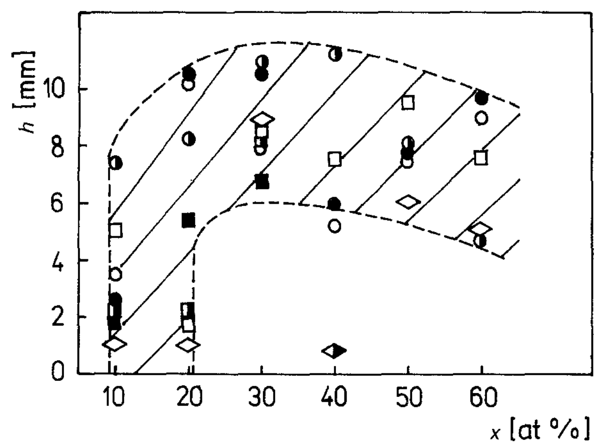

Figure 2 Concentration dependence of average height h of oblique-mode fracture for $Fe_{80-x}Ni_xB_{20}$ alloys. 300 K = circles; 77 K = squares; 4.2 K = rhombus; $3.3 \cdot 10^{-6}\,s^{-1}$ = full symbols; $1.25 \cdot 10^{-4}\,s^{-1}$ = empty symbols; $1.25 \cdot 10^{-3}\,s^{-1}$ = half-full symbols.

TABLE II The average values of α and 90% intervals of reliability for different Ni atoms concentrations, strain rates and temperatures. x — Ni atoms concentration, $\dot{\varepsilon}$ — strain rate, $\bar{y}_x$ — column average, $\bar{y}_{\dot{\varepsilon}}$ — line average.

	300 K				
x	$\dot{\varepsilon}$ 3.3 × 10^{-6}	1.33 × 10^{-5}	1.25 × 10^{-4}	1.25 × 10^{-3}	$\bar{y}_{\dot{\varepsilon}}$
10	45.0 ± 6.5	39.6 ± 4.1	59.4 ± 3.5	40.0 ± 4.6	45.9
20	42.7 ± 5.3	43.3 ± 3.8	38.7 ± 3.8	40.4 ± 3.3	41.6
30	40.0 ± 3.8	36.2 ± 4.1	34.3 ± 3.5	39.0 ± 2.8	37.4
40	42.5 ± 4.6	38.3 ± 5.3	43.8 ± 4.6	39.3 ± 3.5	40.8
$\bar{y}_x$	42.5	39.8	43.6	39.7	
x	77 K				
10	–	–	63.8 ± 3.2	64.3 ± 3.6	64.1
20	–	–	65.0 ± 2.8	60.0 ± 4.5	62.5
30	48.8 ± 2.8	43.2 ± 2.6	38.4 ± 2.8	44.1 ± 2.4	43.6
40	–	–	48.3 ± 3.6	51.0 ± 2.8	49.7
$\bar{y}_x$	48.8	43.2	53.8	54.9	

of reliability are also shown on the assumption that the experimental data are normally distributed.

Statistical processing has confirmed that α essentially depends (with 99% probability) on nickel atoms concentration. This result is shown in Fig. 3a and b from which it is clear that if the dependence of α on Ni concentration does exist, then it has a minimum at x = 30 at % Ni. Since the Poisson ratio of $Fe_{80-x}Ni_xB_{20}$, alloys varies monotonously in the range x = 10–40 at % [5], then the possible change of α, connected with the change of ν [4] must be monotonous too. Hence, non-monotonous dependence of α on x is not connected with different compressibility of alloys and different ν values. Moreover, the α dependence on ν, predicted by the theory [4], is considerably weaker than that observed experimentally. The observed $\alpha(x)$ dependence can be explained proceeding from the notion of α dependence on dilatancy. The value of dilatancy for metallic glasses is not known at present. But it may be supposed that in the $Fe_{80-x}Ni_xB_{20}$ system dilatancy is minimum for $Fe_{50}Ni_{30}B_{20}$ on the ground of some thermodynamic characteristics of this system.

It is known [9] that $Fe_{50}Ni_{30}B_{20}$ is the most thermodynamically stable alloy. It has maximum crystallization temperature T_c and minimum crystallization heat ΔH_c in the system studied (Fig. 4). If we consider ΔH_c using the configurational approach, then its minimal value must correspond to the minimal change of specific volume during crystallization. Since some researchers suppose [10] that the atomic rearrangements during crystallization and plastic deformation of glasses are similar, it is natural to expect that the specific volume change ΔV at plastic deformation is

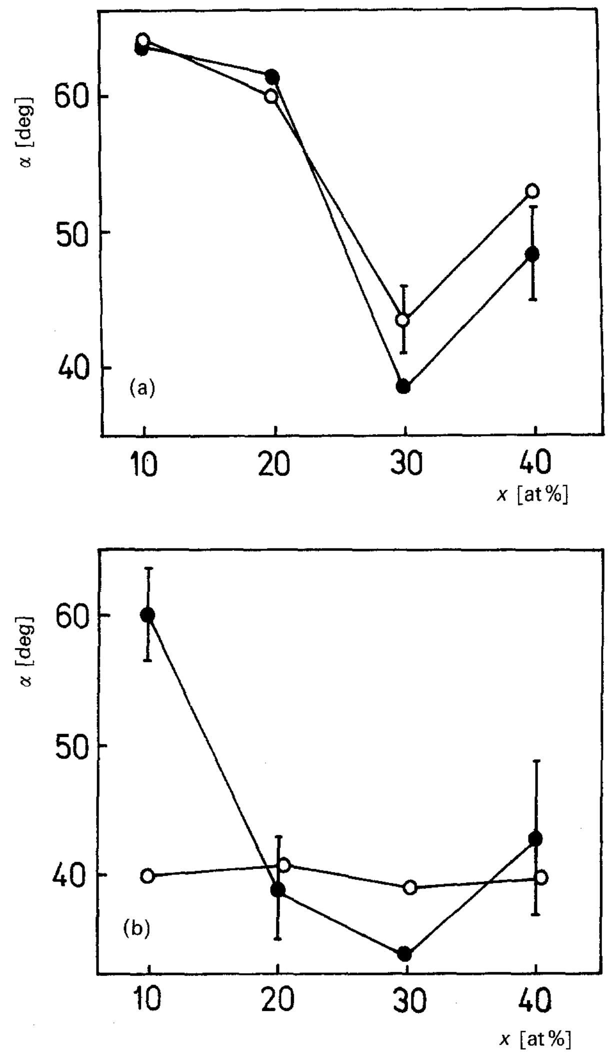

Figure 3 Concentration dependence of average angle α of oblique mode orientation in $Fe_{80-x}Ni_xB_{20}$ at strain rates $1.25 \cdot 10^{-3}\ s^{-1}$ (○) and $1.25 \cdot 10^{-4}\ s^{-1}$ (●): (a) 77 K, (b) 300 K.

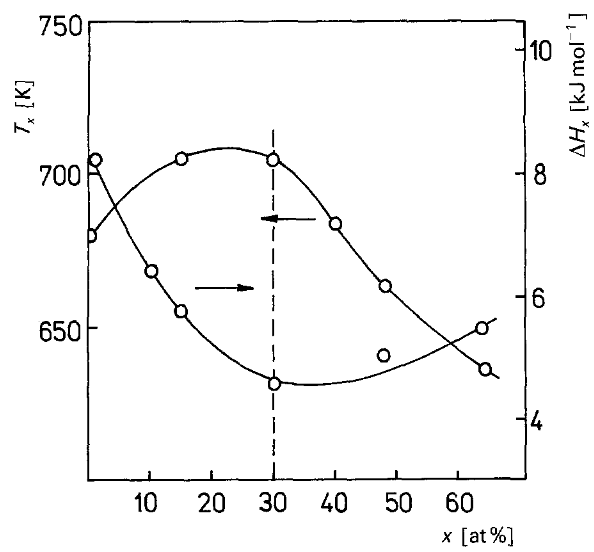

Figure 4 Concentration dependence of crystallization heat and initial crystallization temperature T_x in $Fe_{80-x}Ni_xB_{20}$ [9].

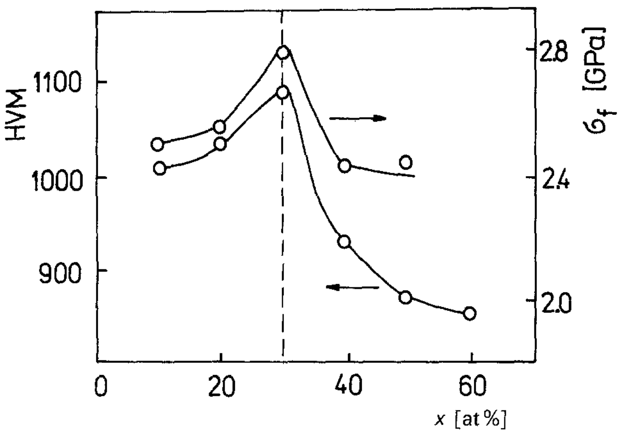

Figure 5 Concentration dependence of fracture stress σ_f and microhardness HVM in $Fe_{80-x}Ni_xB_{20}$ [11].

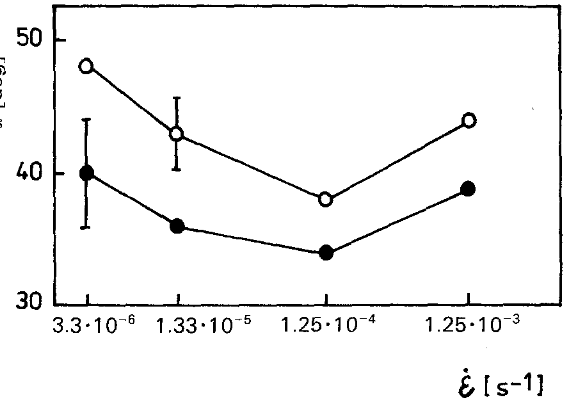

Figure 7 Strain rate dependence of the average angle α of oblique mode orientation in $Fe_{50}Ni_{30}B_{20}$ at 77 K (○) and 300 K (●).

also minimal for $Fe_{50}Ni_{30}B_{20}$. Then dilatancy of this alloy is minimum as well, which explains the discovered minimum of α. Such an idea could be checked indirectly by measuring the hydrostatic pressure effect on the flow stress of $Fe_{80-x}Ni_xB_{20}$. However, such data are not known to the authors.

The concept of minimum dilatancy expected for $Fe_{50}Ni_{30}B_{20}$ also agrees with the data on the strength of MGR $Fe_{80-x}Ni_xB_{20}$. It is known that in this system non-monotonous concentration dependence of microhardness *HVM* and fracture stress σ_f at ductile shear fracture under tension is observed [11] (Fig. 5). Maximum σ_f and *HVM* are reached for $Fe_{50}Ni_{30}B_{20}$, where possible splitting of material owing to dilatancy under nucleation and development of plastic shear is minimal. Figure 6 showing the correlation field that characterizes the relation between h and σ_f serves as an evidence of the dilatancy y-σ_f correlation existence. Unequivocal clarification of the physical nature of this phenomenon requires further development of the theory of plastic deformation and fracture of metallic glasses.

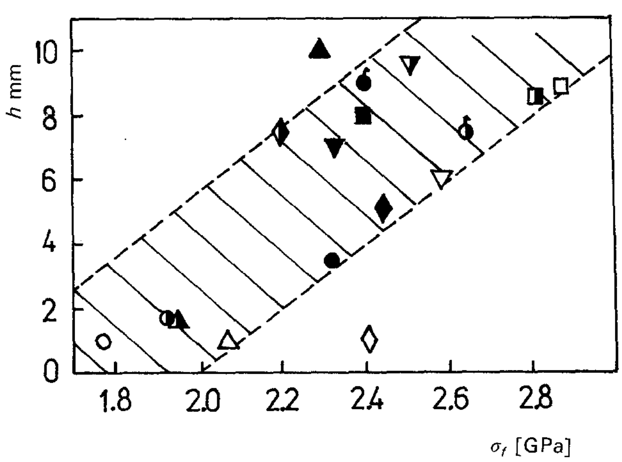

Figure 6 Correlation of the average height h of oblique fracture mode and fracture stress σ_f of $Fe_{80-x}Ni_xB_{20}$ at the strain rate $1.25\cdot10^{-4}\,s^{-1}$. 300 K – full symbols; 77 K – half-full symbols; 4.2 K – empty symbols. $x = 10$ – circles; $x = 20$ – triangles; $x = 30$ – squares; $x = 40$ – rhombs; $x = 50$ – overturned triangles; $x = 60$ – circles with strokes.

3.3. Strain rate and temperature effect on oblique mode orientation

Statistical analysis of Table II data shows that though the strain rate dependence of α is significantly weaker than concentrational one, the strain rate effect is so essential that the α on x dependence vanishes under some strain rates and temperatures (Fig. 3b). This may indicate that thermal activation and relaxation processes affect nucleation and development of plastic shear followed by fracture.

Note that such effect is not equivocal. On the one hand, free volume relaxation near the elementary shear must cause dilatancy decrease with the decrease in $\dot{\varepsilon}$. On the other hand, the anticipated increase of the elementary shears number with the decrease in $\dot{\varepsilon}$ [12] may, in its turn, increase the dilatancy. Figure 7 demonstrates that the influence of $\dot{\varepsilon}$ on α is complex indeed. It is seen that at the given x there is a minimum α that must correspond to minimal dilatancy. However, it was established [12] that at the same strain rate there exists a minimum of σ_f, and the concentration dependence of σ_f is weak. Hence, the supposed correlation of dilatancy and σ_f would have been broken. These inconsistencies emphasize the necessity of a preliminary theoretical analysis, taking into account the elementary shears interactions [13].

The temperature effect on α manifests itself in a simple increase of α with the decrease of temperature, non-monotonous dependence $\alpha(x)$ being conserved (Fig. 7).

3.4. Mean amplitude of quenching stresses in MGR

The spread of experimental values of α for each alloy may be due to internal quenching stresses in ribbons or to a combination of oblique and "45-degree" modes (the trace of the latter on ribbon surface is orthogonal to tension axis). But systematical deviations of measured α to lower and not higher values than theoretical ones [3, 4] make us assume that these deviations are not due to the "45-degree" mode, but to quenching stresses. Longitudinal bending of MGR under investigation (Fig. 8) is a macroscopic manifestation of these stresses.

Knowing the sag y of the ribbon with the width w, it is possible to estimate the difference $\Delta\sigma_q$ of the

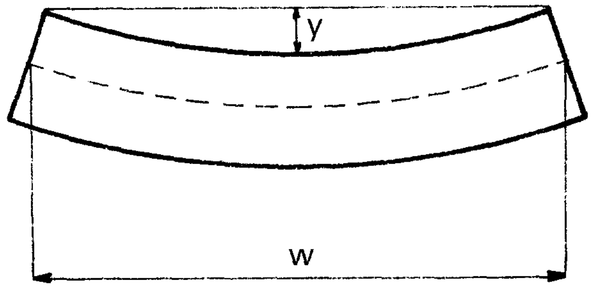

Figure 8 Schematic view of ribbon cross-section under longitudinal bend.

quenching stress σ_q on the contact (dull) and noncontact (shining) surfaces of a ribbon:

$$\Delta\sigma_q = \frac{4ty}{w^2} E \quad (1)$$

where t is the ribbon thickness, E is the Young modulus. Measured values of y ranged from 150 to 210 μm, that gives $\Delta\sigma_q = 25 \div 35$ MPa.

Since $\Delta\sigma_q$ is oriented perpendicularly to the ribbons axis, it is natural to assume the same orientation for σ_q. Then under tension the principal stress axis becomes inclined to that of the ribbon by the angle $\Delta\alpha$:

$$\Delta\alpha = arctg \frac{\sigma_q}{\sigma} \quad (2)$$

with σ being the tension stress. In this case

$$\alpha = \alpha_{th} - \Delta\alpha \quad (3)$$

α_{th} being the angle, predicted by [3, 4]. This explains the sign of the α deviation from α_{th}. Since the absolute values of $\Delta\alpha$ are as high as 8°, then for $\sigma_q = \sigma_f \, tg \, \Delta\alpha$ we obtain $\sigma_q \cong 0.14 \ \sigma_f \cong 420$ MPa. It is in good agreement with the literature data [14].

Hence, it follows that the existence of internal quenching stresses in MGR may be responsible for the main difficulties of dilatancy estimation. It seems that the unreal dilatancy estimate in [8] was caused by disregarding the possible internal stresses. Since the concentration dependence of σ_q is not known, the concentration dependence of the dilatancy supposed above must be regarded carefully.

Thus, the observed difference of oblique mode failure crack orientations from the theoretical ones can be explained by a kind of elastic anisotropy due to quenching stresses, mainly stemming from non-uniform heat removal during the MGR preparation.

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